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Mechanical Properties of a Metal-Matrix Composite Based on Copper and Aluminum, Obtained via Shear Deformation under Pressure

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Abstract—Results are presented from studying the structure and mechanical properties of an Al–Cu–Al metal-matrix composite obtained via shear under pressure on Bridgman anvils with grooves. The tensile strength is 485 MPa, considerably higher than that of either pure aluminum or copper. The main mechanism of failure is a viscous fracture along the Al matrix with no notable stratification along the interphase boundaries.

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INTRODUCTION

In recent years, the attention of many researchers has been riveted on metal-matrix composites (MMCs) based on aluminum, due to their high values of specific strength, hardness, wear resistance, and fatigue durability [1–5]. Composites strengthened by introducing dispersive particles are usually characterized by weak bonds between the particles and the matrix and nonuniform distributions of the reinforcing particles. The solid solution strengthening of aluminum is limited by the maximum solubility of the components. It is known that reactions can be stimulated not only by elevated temperature but by intense plastic deformation as well [6]. Ways of obtaining metal-matrix composites via deformation have thus been developed in Russia and abroad in recent years. There are a number of approaches to intense plastic deformation for obtaining compounds from different metals: diffusion welding, the mechanical alloying of powders, pack rolling, and explosion welding [7, 8]. In contrast to these, shear deformation under pressure allows us not only to create nanostructure but to join metals that do not form compounds according to a state diagram. Over a relatively short period of time, we can obtain in one step monolithic samples of sufficient size for determining structure and physical and chemical properties even at room temperature [9]. For example, we have used shear deformation under pressure to manufacture monolithic composites from initial layered components of pure Al and Cu, with subsequent analysis of their structure [10–12]. This work is a continuation of our earlier studies and presents results from investigating the mechanical properties of a metal-matrix Al–Cu composite, obtained via shear deformation under pressure.

EXPERIMENTAL

Pure Al (99.3–99.5 wt %) and Cu (99.90 wt %) were used as the initial materials for obtaining our composite. Blanks of coarse-grained aluminum and copper were obtained by annealing the initial materials at 400 and 900°C, respectively, in order to obtain a homogeneous structure.

Our disks were made of rods. Samples of the metal matrix composites were obtained via shear deformation under pressure on Bridgman anvils with grooves 12 mm in diameter and 0.25 mm deep at 5 GPa, and 10 revolutions at a rate of 1 rpm at room temperature. Prior to deformation, the surface of the disk was reduced and stacked by interleaving layers of Al–Cu–Al. Figure 1a shows the pattern of blank packing. The configuration of one composite after deformation is presented in Fig. 1b.

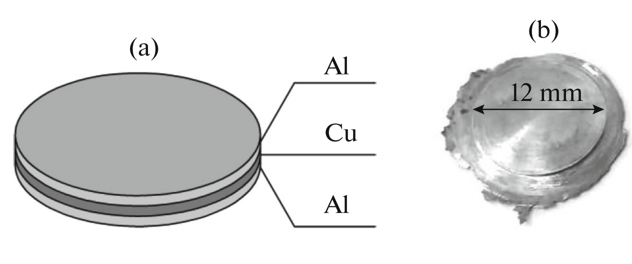

Fig. 1. (a) Scheme of Al–Cu–Al blank packing and (b) general view of the Al–Cu–Al composite sample obtained via shear deformation under pressure.

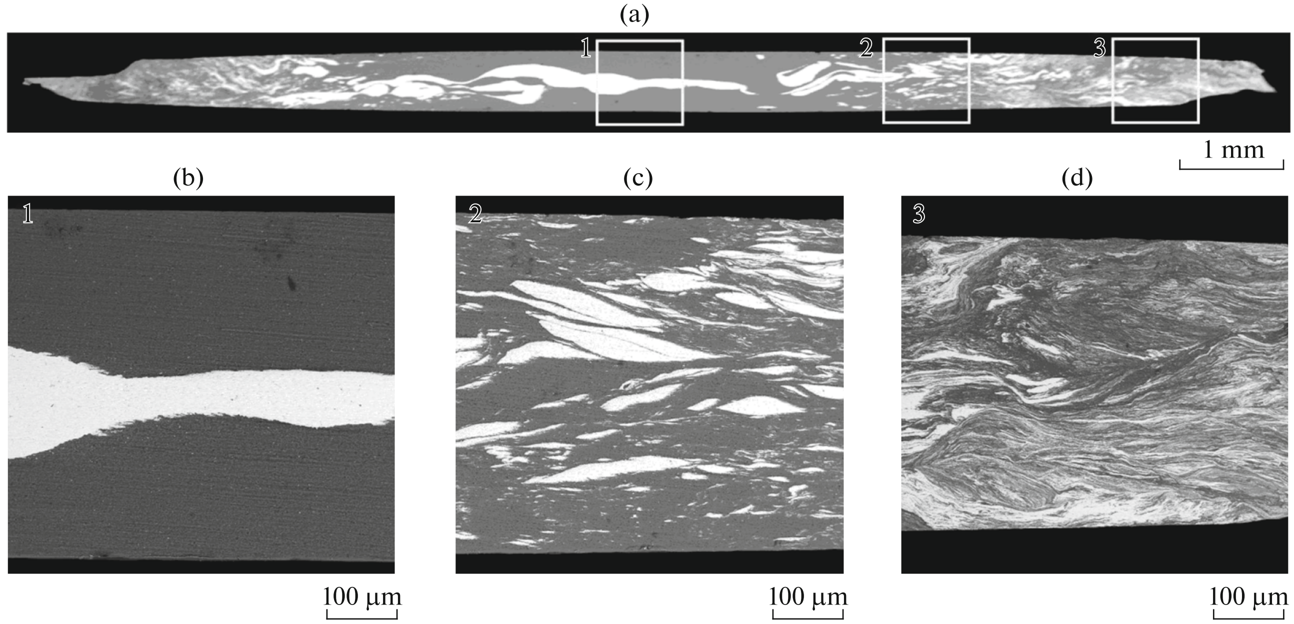

Fig. 2. (a) Microstructure (BSE mode) of disc cross sections obtained for the Al–Cu–Al composite at low magnification; (b) magnified images of the zones in the center, (c) in the middle of the radius, and (d) near the edge [7].

As was shown in [10–12] via electron microscope analysis of the cross section of a deformed Al–Cu–Al composite, the sample was monolithic and had no pores. However, the mixing of the components is seen to vary, depending on the distance from the sample's center (Fig. 2).

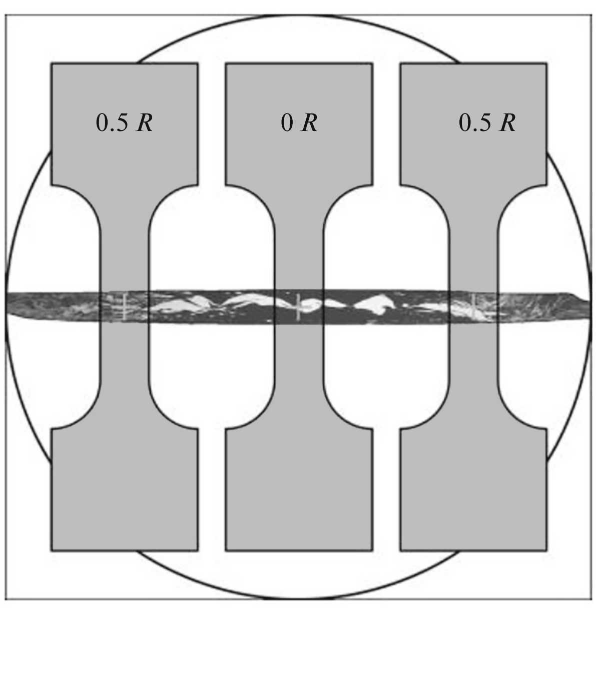

Fig. 3. Cutting pattern for the tensile samples.

The best mixing and more homogeneous structures are observed in the middle of the radius (Fig. 2c). The samples used for structural tensile tests were therefore cut out at the middle of the radius. Test samples were also cut from the central zone with bad mixing, for purposes of comparison. Three samples were thus cut from each composite disk for mechanical testing. The pattern of disk layering and the microstructure in the working zone of the strain samples are shown in Fig. 3.

Annealing was done to obtain composites with inclusions of intermetallic phases. Annealing temperatures of 350 and 450°C were selected on the basis of DS measurements (which showed a wide peak in the interval of 150–450°C) and X-ray structural analysis (which showed the formation of a substantial share of Al_2Cu, AlCu, Cu_9Al_4 intermetallics after annealing at 350 and 450°C). Samples for investigating mechanical properties according to the scheme described above were cut out after thermal treatment. Rupture tests were performed with specially developed tooling (Fig. 4) on an Instron Model 1185 dynamometer. The use of shear deformation under pressure on anvils 12 mm in diameter with grooves 0.5mm deep allowed tests of the strain on commensurate samples corresponding to GOST 1497-84. Destructive tests were performed at a tension rate of 1 mm/min, with registration of the load–transition (P–Δl) diagram. Initial diagrams were subsequently developed. The composite's plasticity was not determined, due to the small working base of the samples.

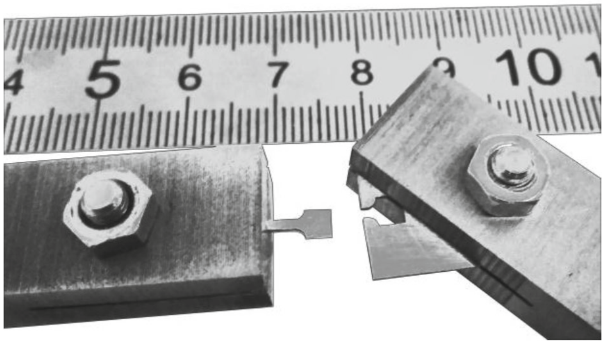

Fig. 4. Tooling for tensile samples.

RESULTS AND DISCUSSION

Figure 5 shows the dependence diagrams of the load–displacement crossbeam during tensile testing of the metal matrix composite obtained via shear deformation under pressure with and without 15 min of treatment at 450°C.

Tensile testing of the metal-matrix composite at room temperature showed that the average value of the ultimate tensile strength of off-centered samples (Fig. 6) was as high as 485 MPa. This exceeds by several hundred percent the ultimate tensile strength of the initial components obtained under similar conditions at the same level and rate of deformation in Bridgman anvils with grooves, and pure Al (~80 Mpa) and Cu (~190 Mpa) obtained in the same manner. Samples with working parts cut from the central zone of the composite were weaker; their average strength was 50% lower, at 245 MPa. Annealing the composite at 150°C substantially reduced the strength in the central part of the composite and at the middle of the radius. Raising the temperature of annealing to 450°C further reduced the strength (Fig. 6) of the samples from the central part and those cut from the central zone with displacement.

The high values of the composite's tensile strength are associated with dispersion of the structure of the aluminum layers and the formation of dispersive and more durable layers of copper in the aluminum matrix, due to the mixing of the components during shear deformation under pressure. The drop in tensile strength as the temperature of annealing rises is due to the emission of intermetallic particles at the Al/Cu interphase boundaries, which makes the material more brittle.

Figure 7 shows images of the surfaces of fractures in composite samples without annealing, with the working part near the middle of the radius (Fig. 7a) and in the central zone of the composite disc (Fig. 7b). In both samples, the character of the fracture is mixed and a ductile fracture is observed in the Al matrix; it is seen as the uniform pattern of pits at high magnification, while Cu sections had fractures of the brittle type. Al and Cu were identified via energy-dispersive spectroscopy (EDS). In the sample cut from the middle of the radius, the sections with the ductile fracture of the Al component alternate with small, uniformly distributed sections of the Cu component, where the failure is accompanied by splitting, both inside the Cu component and at the boundary of the junction with Al (Fig. 7a). The structure of the sample as a whole is uniform, and analysis of the fracture shows that the interphase boundaries of the Al/Cu compound facilitated the dissipation of destructive energy, due to the deviation and branching of cracks along the Al/Cu interface and the ductile fracture of the Al matrix. This ensured the maximum strength of the sample.

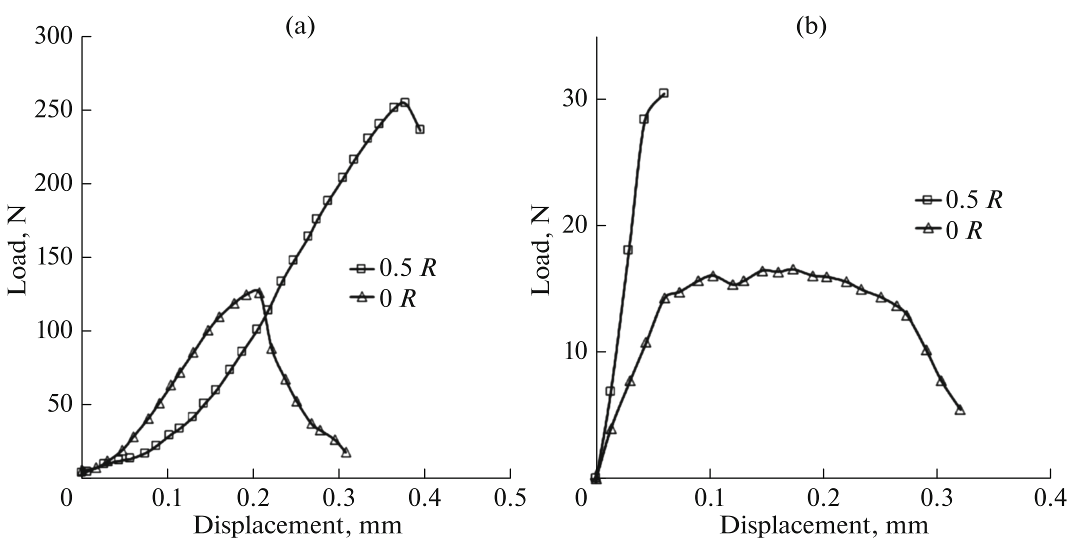

Fig. 5. Strength diagrams of the fracturing in the metal-matrix composite samples: (a) without thermal treatment, (b) with 15 min of thermal treatment at 450°C.

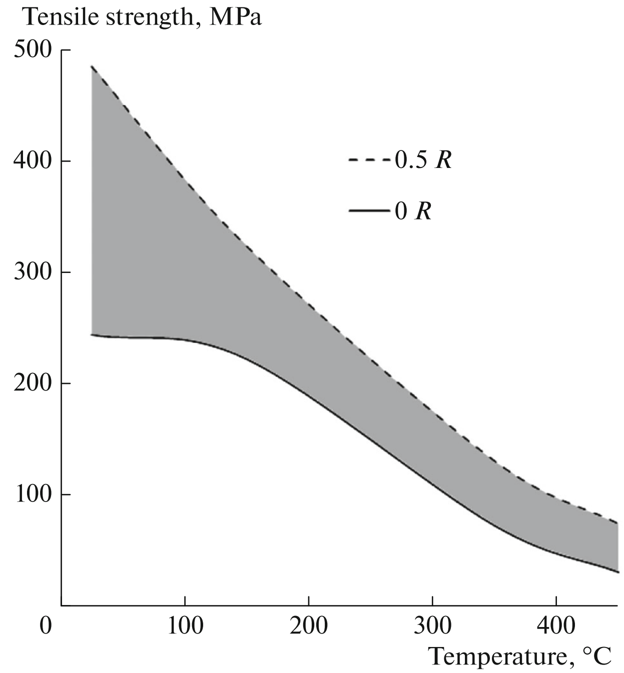

Fig. 6. Dependence of the strength of the metal-matrix Cu–Al composite on the temperature of annealing.

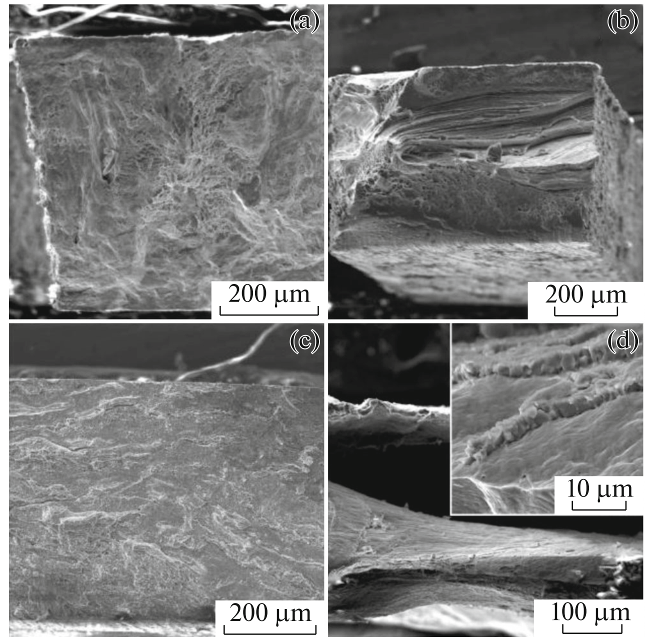

Fig. 7. Fractography of the Al–Cu–Al composite samples (a, b) without annealing and (c, d) after annealing at 450°C with the working part (a, c) near the middle of the radius and (b, c) in the central zone of the composite disc.

In the sample cut from the central zone (Fig. 7b), where the mixing of the Al and Cu components was negligible, a fracture was observed along the boundary of the contiguous metals. Splitting along planes was observed in Cu, mostly in parallel to the surface of the composite (Fig. 7b). Pits observed on the Al phase testified to its ductile fracture. The surface of the fracture in the copper component was deeper than the one in the Al component in both parts of the sample, indicating that the copper component fractured first.

Annealing the composite at 150°C reduced the viscosity of the Al component in the sample cut from the middle of the radius, and the number of sections with pit fractures fell. However, no appreciable difference was observed between the fractures in the samples annealed at 15°C and those without annealing. The small drop in tensile strength after annealing was probably due to the onset of the decay of the Al solid solution that formed at the Al and Cu interface during shear deformation under pressure, and the formation of Guinier–Preston zones, whose small size does not allow their differentiation on fractures. As with the sample without annealing, the fracture in the sample from the central zone occurred at the Al and Cu interface. In the Al component, the fracture proceeded according to a ductile mechanism with the formation of pits, and the tensile strength before and after annealing was virtually the same (Fig. 6).

After annealing at 350°C the pattern of fractures changed substantially. The fracture in the sample with the working part near the middle of the radius was brittle. Clearly faceted grains of intermetallic inclusions inside the matrix of the Al phase were seen along with the Al and Cu, and the matrix remained viscous, as can be seen from the pits and twisting lines of the fracture. Intermetallic grains are in this case seen inside the pits at high magnification. In general, however, even the small number of intermetallic particles at the interphase Al/Cu boundaries in this sample resulted in an almost 300% reduction in tensile strength (Fig. 6).

After annealing at 450°C (Figs. 7c, 7d) there was coarsening of the intermetallic particles and an increase in their volume ratio, as is clear from the data of our X-ray phase analysis. This led to fracturing of the samples cut from the middle of the radius and outside the central zone along the grain boundaries of the intermetallic particles (i.e., intercrystalline fracturing).

CONCLUSIONS

Strain deformation of sandwich-type Al–Cu–Al composite samples, obtained via shear deformation of our initial coarse-grained plates of aluminum and copper under pressure, demonstrated the strong dependence of tensile strength on the level of component mixing. In the central zone, where mixing was weak, the tensile strength was half that of the zone in the middle of the radius, where uniform mixing of the components was observed during shear deformation under pressure. The tensile strength was 485 MPa, considerably higher than that of either pure aluminum or copper. The main mechanism during tensile testing was a ductile fracture along the Al matrix with no notable stratification along interphase boundaries. In the samples with bad mixing, fracturing was mainly due to stratification along the Al/Cu interfaces. Annealing at 150, 350, and 450°C lowered the tensile strength of the samples, due to the phase transformations and the emission of intermetallic phases at the Al/Cu interfaces. There was fracturing in the annealed samples of the composites, due to brittle splitting along intermetallic grains at the interphase boundaries.

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